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AND ITS INFLUENCE ON THE STRESS-STRAIN BEHAVIOR AT
VERY HIGH STRAIN RATES

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The Rate Dependence of Structure Evolution in Copper and its Influence on the Stress-Strain Behavior at Very High Strain Rates

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Introduction

A constitutive description of the deformation copper based on the internal-state-variable model of Kocks (1) and Mecking and Kocks (2) has recently been proposed (3). This model uses the mechanical threshold stress, or flow stress at 0 K, as an internal-state variable that represents a mechanical measure of the microstructural state. Another feature of the model is that evolution of the internal-state variable (structure evolution) is treated separately from the determination of the strain-rate and temperature dependent flow stress for any microstructural state. This treatment was motivated by the observation in polycrystalline copper that the yield stress at any standard strain rate and temperature (e.g., $\dot{\epsilon}=10^{-3} \text{ s}^{-1}$ and 295 K) on samples prestrained at room temperature to a constant strain varied with the strain rate of the prestrain process. This is illustrated in Fig. 1 which shows the stress-strain histories for specimens deformed at $\dot{\epsilon}=1.4 \times 10^{-3} \text{ s}^{-1}$ and $\dot{\epsilon}=10^{-4} \text{ s}^{-1}$ to a strain of $\epsilon=.15$ and reloaded at $\dot{\epsilon}=10^{-3} \text{ s}^{-1}$ and 295 K. The reload yield stress on the sample deformed at the high strain rate exceeds that for the sample deformed at the low strain rate by 45 MPa. These data illustrate the strain-rate history effect that has been reviewed recently by Klepaczko and Chiem (4). Figure 2 shows the variation of the reload yield stress with strain rate for samples deformed again to $\epsilon=.15$ at strain rates from 10^{-4} s^{-1} to 10^{-1} s^{-1} . It is evident in this figure that at strain rates exceeding 10^{-3} s^{-1} the strain-rate history effect becomes more pronounced; that is, the strain-rate sensitivity, measured at constant strain in this plot, increases dramatically at high strain rates. This is the same increased strain-rate sensitivity that is found in a plot of flow stress at the prestrain strain rate (rather than at the reload strain rate) versus strain rate. These data indicate that the increased strain-rate sensitivity found at these strain rates is not due to a change in the rate controlling deformation mechanism but rather is a strain-rate history effect.

In previous studies we have quantified these observations (5) and have fit the behavior to the Kocks/Mecking internal-state-variable constitutive model. The purpose of the work reported here is to add the influence of dislocation drag (phonon drag) at high strain rates and to extrapolate the model beyond the strain-rate regime to which it has been fit.

The Model

For any microstructural state, represented by the mechanical threshold stress $\hat{\sigma}$, deformation is assumed to be controlled by the thermally activated interactions between mobile dislocations and forest dislocations and to be described by a phenomenological law of the form (3)

$$\sigma = \hat{\sigma}_a + (\hat{\sigma} - \sigma_a) \left[1 - \frac{kT \log(\dot{\epsilon}_0/\dot{\epsilon})}{\hat{\sigma}_0 \mu b^3} \right]^{3/2} \quad (1)$$

where σ is the applied stress, σ_a is an athermal stress representing the long-range stress field (e.g., with grain boundaries, etc.), k is the Boltzmann constant, T is the temperature, $\dot{\epsilon}_0$ is a constant, $\dot{\epsilon}$ is the strain rate, $\hat{\sigma}_0$ is the normalized total activation energy, μ is the polycrystalline shear modulus and b is the Burgers vector. The constants σ_a , $\dot{\epsilon}_0$, and $\hat{\sigma}_0$ (40 MPa, 10^{-3} s^{-1} , and 1.6, respectively) are assumed not to vary.

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Evolution of the structure parameter $\hat{\sigma}$ is considered as the balance between dislocation accumulation and dynamic recovery and the strain-hardening rate $\theta = d\hat{\sigma}/d\epsilon$ is used to characterize the differential variation of the structure parameter with strain. We assume that the structure will eventually saturate at a strain-rate and temperature dependent stress σ_s , leading to a zero rate of strain hardening and we model the strain-hardening behavior using an equation of the form

$$d\hat{\sigma}/d\epsilon = \theta = \theta_0 \left[1 - F \left(\frac{\hat{\sigma}^2 - \sigma_0^2}{\sigma_s^2 - \sigma_0^2} \right) \right] \quad (2)$$

where θ_0 is the strain-hardening due to dislocation accumulation and the factor F is chosen from experimental results to be

$$F = \frac{\tanh(2x)}{\tanh(2)} \quad (3)$$

where

$$x = \left(\frac{\hat{\sigma} - \sigma_0}{\sigma_s - \sigma_0} \right)^2.$$

This choice for F is made solely to fit the experimental results and we do not imply any physical significance to this form. However, this F is only slightly different from $F=1$, which represents Voce law behavior (1). The strain-rate and temperature dependence of the saturation stress σ_s is introduced using a phenomenological cross-slip model (1,6) which is written

$$\log(\dot{\epsilon} / \dot{\epsilon}_{s0}) = \frac{u b^3 A}{k T} \log(\sigma_s / \sigma_{sc}) \quad (4)$$

where $\dot{\epsilon}_{s0}$, A , and σ_{sc} are constants ($6.2 \times 10^{10} \text{ s}^{-1}$, 0.312 , and 900 MPA , respectively).

One significant difference between the form of Equ. 2 found to describe the copper results over a wide range of strain rates and that used earlier over a limited range of strain rates (1,2) is that at high strain rates the data fits indicate a strain-rate dependence to the θ_0 term, given by

$$\theta_0 (\text{MPa}) = 2390 + 12 \log \dot{\epsilon} + 0.037 \dot{\epsilon}. \quad (5)$$

The linear term in this expression begins to contribute significantly at strain rates exceeding $3 \times 10^5 \text{ s}^{-1}$ which yields the strong history dependence or constant-strain, strain-rate sensitivity found at these high strain rates. Clearly, Equ. 5 can not be valid at strain rates greater than 10^6 s^{-1} because the θ_0 term rises above the shear modulus at this strain rate. Thus for predictions at strain rates greater than this, the maximum θ_0 value will be restricted to the shear modulus. We do not understand the origin of the enhanced dislocation generation rate but speculated in (3) that the dislocation storage distance might become dislocation velocity, or phonon drag, limited at these strain rates. The addition of phonon drag controlled deformation to the model, which is the subject of the next section, appears to shed further light on this issue.

Dislocation Drag Controlled Deformation

The procedure to combine the kinetics of dislocation drag controlled deformation with those for thermally activated controlled deformation has been described previously. (7,8,9) Generally these analyses apply the well known relation between strain rate, mobile dislocation density ρ_m , and average dislocation velocity V , which is written

$$\dot{\epsilon} = \alpha M \rho_m b V = \alpha M \rho_m b \frac{\lambda}{\tau_r + \tau_w} = \frac{\dot{\epsilon}_0}{\tau_r \tau_0 + \exp\left(\frac{\Delta G}{k T}\right)} \quad (6)$$

where α is a constant of order 0.5, λ is the mean distance between obstacles, t_w is the time spent waiting for thermal energy to assist a dislocation past an obstacle, t_r is the time spent running to the next obstacle, ν_o is the jump or attempt frequency (10^{11} s^{-1}), and ΔG is the activation energy, which can be deduced from Equ. 1. The running time is the mean distance between obstacles divided by the drag controlled velocity, which yields

$$\dot{\epsilon} = \frac{\dot{\epsilon}_0}{\frac{MB\lambda\nu_o}{\sigma b} + \exp\left(\frac{\Delta G}{kT}\right)} \quad (7)$$

where M is a Taylor factor and B is the drag coefficient. To restrict the dislocation velocity to values less than the shear wave velocity ($C_s = (\mu/\rho)^{1/2} = 2170 \text{ m/s}$ at 295 K) we replace the constant drag coefficient with an effective drag coefficient,

$$B = \frac{B_o}{\left[1 - \left(\frac{v}{C_s}\right)\right]^{1/2}} = \left[B_o + \left(\frac{b\sigma}{C_s M}\right)^2\right]^{1/2} \quad (8)$$

where B_o is the drag coefficient deduced from ultrasonic or etch pit techniques ($B_o = 4.3 \times 10^{-5} \text{ Pa s}$). The value of λ is the only remaining unknown and for this we will investigate two possibilities; the first assumes that λ is a constant while for the second we assume that

$$\lambda = \frac{1}{\sqrt{\rho}} = \frac{\alpha \mu b}{\sigma} \quad (9)$$

1. λ Constant

Equation 7 with Equ. 8 can be substituted for Equ. 1 and solved for the flow stress for any combination of strain rate, temperature, and mechanical threshold stress. Predictions for the variation of the flow stress with strain rate as a function of strain are shown in Fig. 3 for $\lambda = 1 \mu\text{m}$ and $\lambda = 0.1 \mu\text{m}$. The gradual increase in strain-rate sensitivity found at $\dot{\epsilon} = 10^4 \text{ s}^{-1}$ is due to the rapid evolution predicted in Equ. 5. The dramatic increase in strain-rate sensitivity found at roughly $\dot{\epsilon} = 10^5 \text{ s}^{-1}$ for $\lambda = 1 \mu\text{m}$ and $\dot{\epsilon} = 10^6 \text{ s}^{-1}$ for $\lambda = 0.1 \mu\text{m}$ is due to the contribution of dislocation drag to the deformation kinetics. As expected, the strain rate where dislocation drag begins to limit deformation is inversely related to the assumed average spacing between obstacles.

2. λ Determined from Equ. 9

A more realistic treatment is to assume that the average spacing between obstacles is inversely related to the square-root of the total dislocation density, as given in Equ. 9. This allows for the value of λ to vary with strain as well as with strain rate and temperature. The predicted variation of the flow stress at $\epsilon = 0.10$ with strain rate is shown in Fig. 4. The different regimes of behavior are identified in this figure and the trends in absence of a relativistic limit to the dislocation velocity and in absence of dislocation drag effects completely also are illustrated. At strain rates exceeding $\dot{\epsilon} = 10^6 \text{ s}^{-1}$ the strain-rate sensitivity is predicted to abruptly decrease as the strain-hardening rate due to dislocation accumulation is restricted to the value of the shear modulus. Figure 5 gives the predicted behavior as a function of strain. It is evident that an actual yield drop is predicted. This is more clearly shown in Fig. 6 which gives stress-strain curves for strain rates from 1 to 10^6 s^{-1} . The yield drop found at strain rates of $\dot{\epsilon} \geq 10^3 \text{ s}^{-1}$ is due to the low total dislocation density and the high value of λ predicted initially. Deformation at low strains and high strain rates, thus, is in the dislocation drag controlled regime which leads to the high initial stress levels.

Discussion

The behavior shown in Figs. 3-6 at strain rates less than 10^4 s^{-1} represents a

fit of the modeling procedure outlined earlier to an extensive series of data. At higher strain rates, the predicted behavior is found by extrapolating the model beyond the regime to which it has been fit. Such extrapolations are valid only if the physical deformation processes that have been modeled remain the same. Although experimental results at the higher strain rates required to test the predictions are difficult to obtain, Fig. 7 shows one comparison with two results of Clifton on similar material at a strain rate of $5 \times 10^5 \text{ s}^{-1}$. (10) The predicted behavior at very low strains differs markedly from that measured, but it is unclear whether a yield drop such as that predicted could be resolved in this experiment. The deviations between experiment and predicted behavior at larger strains are not serious when the difference in stress state between the pressure/shear experiment and the uniaxial compression experimental results used to fit the model is accounted for through a Taylor factor. One feature of the experimental results which is predicted is the apparent lack of strain hardening at strains exceeding $\epsilon=0.10$. This is a result of the rapid strain-hardening rate at low strains and thus the early saturation of the dislocation substructure. This indicates that at high strain rates the stress-strain curves tend toward perfect plastic behavior. This trend illustrates one of the limits imposed by the physically-based modeling procedure described earlier. Another limit is that imposed by the shear modulus on the dislocation accumulation rate, or $\dot{\theta}_0$, in Equ. 5, which as shown in Figs. 4 and 5, begins to influence the behavior at $\dot{\epsilon}=10^3 \text{ s}^{-1}$.

The curves in Figs. 3-5 indicate a limit to the imposed strain rate that can be sustained by the deformation mechanisms considered in the model. For the curves in Figs. 4 and 5, this limit is roughly $\dot{\epsilon}=10^3 \text{ s}^{-1}$. It is expected that when the stress begins to rise as abruptly as shown in these figures, deformation twinning or localized deformation in the form of shear bands may begin to contribute to the deformation process.

The high stresses found near yield at strain rates exceeding 10^3 s^{-1} were described as resulting from the low initial dislocation density and the influence of dislocation drag on the stress required to move these dislocations at the required high velocities. Yield drops such as those shown in Fig. 6 are more typical of bcc metals, and indeed the explanation of this behavior generally focuses on the low initial mobile dislocation density. Models for the yield region that combine an assumed dislocation generation law and the experimentally measured stress dependence of the dislocation velocity have been previously proposed. (11) Yield drops also have been observed in fcc metals (12) and, although their presence is often argued to be an experimental artifact, experiments by Harding indicated that the measured high yield stresses and subsequent yield drops were a true material response to the imposed dynamic loading conditions. (12)

The significance of the predicted yield behavior shown in Fig. 6 is that, the high yield stresses in copper are not generated until strain rates of 10^3 s^{-1} are exceeded. This correlates well with the increased dislocation accumulation rate expressed by Equ. 5 which was found to explain the increased strain-rate sensitivity at high strain rates. Thus, whereas we speculated previously (3) about the mechanistic processes that could lead to the observed rapid dislocation generation rates, the predictions shown in Fig. 6 suggest that the high initial stresses may directly be responsible for this behavior. The stress to operate a dislocation source, for instance, is related to the dislocation line tension T and the spacing between obstacles L by

$$\sigma = \frac{T}{b L} \quad (10)$$

From this relation, one could predict that the high initial stress levels at high strain rates would activate sources that would have remained inactive at lower strain rates, which could explain the observed rapid dislocation accumulation rate. The model presented earlier, and in particular the expression given for the dislocation accumulation rate in Equ. 5, does not account for the stress dependence of θ_0 . This likely would be difficult to do without a detailed understanding of the distribution of loop lengths that could be activated and the evolution of this distribution with strain.

Conclusions

A model for the deformation of copper over a wide range of strain rates has been modified to account for the contribution of dislocation drag at high strain rates. The main effect of adding this mechanism is found in the near yield region at strain rates exceeding 10^3 s^{-1} . However, even at strain rates as high as 10^6 s^{-1} the rate controlling deformation mechanism at strains greater than roughly $\epsilon=0.1$ is thermal activation. It is postulated that the high stresses predicted at yield when the strain rates are raised above 10^3 s^{-1} may lead to the rapid dislocation accumulation rates that have been found experimentally and which lead to the increased strain-rate sensitivity reported for copper and other metals at these strain rates.

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References

- (1) U.F. Kocks, ASME J. Engr. Mat. Tech. 98 (1976) 76.
- (2) H. Mecking and U.F. Kocks, Acta Metall. 29 (1981) 1865.
- (3) P.S. Follansbee and U.F. Kocks, Acta Metall. (1987) To appear.
- (4) J.R. Klepaczko and C.Y. Chiem, J. Mech. Phys. Solids 34 (1986) 29.
- (5) P.S. Follansbee, U.F. Kocks, and G. Regazzoni, "DYMAT '85, Mechanical and Physical Behavior of Materials Under Dynamic Loading", Journal De Physique 46 (1985) C5-25.
- (6) P. Haasen, Phil. Mag. 3 (1958) 384.
- (7) P.S. Follansbee, G. Regazzoni and U.F. Kocks, "Mechanical Properties of Materials at High Rates of Strain", Inst. Phys. Conf. Ser. 70, Institute of Physics, London (1984) 71.
- (8) D. Klahn, A.V. Mukherjee and J.E. Dorn, "Strength of Metals and Alloys", ASM (1971), 951.
- (9) R.J. Clifton, "Shock Waves and the Mechanical Properties of Solids" Syracuse Univ. Press, Syracuse, NY, (1971) 73.
- (10) R.J. Clifton, Division of Engineering, Brown University, Providence, RI.
- (11) W.G. Johnston, J. Appl. Phy. 33 (1962) 2716.
- (12) J. Harding, Acta Metall. 19 (1971) 1177.

Figure Captions

Fig. 1: Dynamic and quasi-static stress-strain curves for copper followed by quasi-static reload stress-strain curves.

Fig. 2: Reload yield stress ($\dot{\epsilon}=10^{-3} \text{ s}^{-1}$) versus prestrain strain rate.

Fig. 3: Stress at constant strain versus strain rate for constant λ .

Fig. 4: Stress at $\epsilon=0.10$ versus strain rate for λ given by Equ. 9.

Fig. 5: Stress at constant strain versus strain rate for λ given by Equ. 9.

Fig. 6: Predicted stress-strain curves for various constant strain rates.

Fig. 7: Predicted stress-strain curve at $\dot{\epsilon}=5 \times 10^5 \text{ s}^{-1}$ and comparison with measurements by Clifton.

(1)













